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EFFECT OF PREFERRED ORIENTATION AND RELATED METALLURGICAL PARAMETERS ON MECHANICAL PROPERTIES AND BALLISTIC PERFORMANCE OF HIGH-HARDNESS STEEL ARMOR

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United States Steel Corporation

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June 1974

By G. R. Speich, H. Hu and R. L. Miller

United States Steel Corporation
Research Laboratory
Monroeville, Pennsylvania 15146

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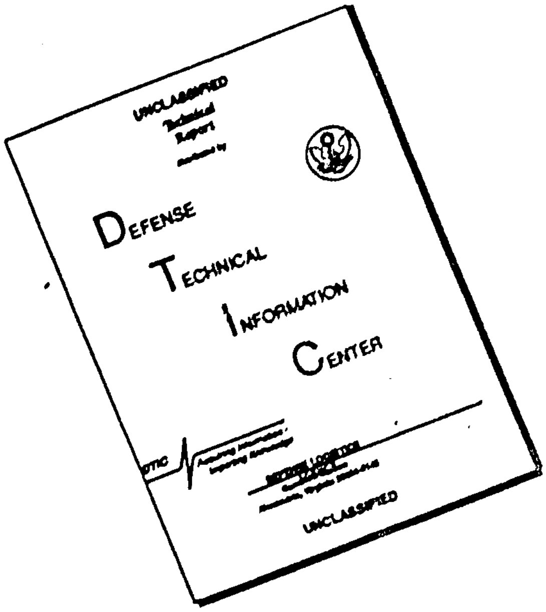

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Optimum ballistic properties were obtained in a medium-carbon 5Ni-Si-Mo-Cu armor steel by rolling at temperatures (∿1500 F) just above the critical range followed by quenching and tempering. Although the mechanical properties of the armor plate were slightly increased by this thermomechanical treatment, these increases were not sufficient to explain the improved ballistic performance. Rather, the improvement also resulted from a simultaneous beneficial change in a number of metallurgical parameters. These

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included a finer martensite plate size and an increase in the amount of retained austenite both of which have a beneficial effect on toughness. Also, the crystallographic texture developed by such low-temperature rolling makes cleavage fracture more difficult both because of a weaker (100) component in the rolling plane of the plate and because of an increase in the elastic modulus in the thickness direction of the plate. Rolling at lower temperatures (1250 F) within the critical range followed by quenching and tempering led to no further increase in ballistic properties.

Because of the low rate of austenite decomposition in this steel, no ferrite formed during intercritical rolling. Thus, thermomechanical treatment performed at temperatures within the critical range was essentially the same as that performed at temperatures just above the critical range.

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This report was prepared by the Research Laboratory of United States Steel Corporation under U. S. Army Contract No. DAAG46-73-C-0244. The contract was administered by the U. S. Army Materials and Mechanics Research Center, Watertown, Massachusetts. This is a final report and covers work conducted from June to December 31, 1973.

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Introduction

Recent work indicated that the ballistic limit of high-hardness steel armor increased if the steel was rolled at a temperature near or slightly below the critical temperature (A_3) followed by quenching and tempering.[1)*] It has been suggested that this beneficial effect might result from the preferred orientation developed in the thermomechanically treated steel armor.[2)]

Preferred orientation will be developed in austenite if it is rolled at such a low temperature that recrystallization is prevented. When the steel is quenched, it transforms into martensite, but this phase transformation only partially destroys the preferred orientation because only a few of the many crystallographic variants are operative in each austenite grain. Thus a texture still remains in the martensitic product.[3)]

The exact mechanism by which this preferred orientation would result in superior ballistic performance of steel armor is still a matter of speculation. Preferred orientation will alter the elastic constants[4)] (elastic wave velocities), and some empirical information suggests that increased elastic constants will improve ballistic performance.[5)] Although increased values of the elastic wave velocities result in a higher stress intensity in the reflected shock wave formed during high-velocity impact,[6)]

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* See References.

the increased elastic constants also make cleavage more difficult. In addition, Bramfitt and Marder[7] have indicated that crystallographic textures which result in a weakening of the (100) component in the rolling plane of steel plate make cleavage in the through-thickness direction more difficult. Since spalling of armor results from cleavage on (100) planes because of the high stress intensity of the reflected tensile shock wave, a weak (100) component should improve ballistic performance.

However, rolling steel armor below its recrystallization temperature may also change the mechanical properties of the plate, alter the austenite grain size and shape, and affect the retained austenite content. All these factors could influence the toughness and ballistic performance of steel armor.

To determine the relationship between the preferred orientation and the related metallurgical parameters developed by thermomechanical heat treatment and the resulting ballistic performance of high-hardness steel armor, a research contract (DAAG 46-73-C-0244) was awarded by the Army Materials and Mechanics Research Center (AMMRC) to United States Steel Corporation.

Materials and Procedure

Two 300-pound melts of armor steel with a composition similar to that of the steel used in our earlier study were prepared in a vacuum induction furnace. The melts were vacuum-cast into rectangular slab ingots 7 by 12 by 24 inches. Their

chemical compositions, along with that of the original steel, are given in Table I. Since all three heats are similar in composition, no differentiation will be made in the text with regard to the particular heat. Most of the experiments were performed with steel from heat No. 7464-8001, although heat No. 7218-8043 was used for some of the initial work.

After melting, the steel ingots were rolled at 2100 F from 7 to 4 inches thick, cut into three equal-sized pieces, and cross-rolled at 2100 F into 2-1/8-inch-thick slabs. These slabs were cut into three smaller slabs, 2-1/8 by 6 by 6 inches. These smaller slabs were given a final thermomechanical treatment consisting of heating to 2100 F, cooling to 1900, 1700, 1500, 1350, or 1250 F, rolling as quickly as possible in 10 consecutive passes to 3/4-inch-thick plate, water quenching after rolling, and tempering 1 hour at 350 F. The temperature of the plates was monitored during rolling by embedding a thermocouple in a hole drilled in the plate. Although the temperature of the plate dropped during rolling at 1900 F (1900 - 1750) and at 1700 F (1700 - 1650), rolling at 1500 and 1365 F closely approximated isothermal conditions because heat generated during the deformation process nearly balanced the heat lost by radiation or convection. Sufficient plate was processed by rolling to produce 5/8- by 12- by 6-inch and 1/2- by 12- by 6-inch ballistic plates and to provide material for metallographic and mechanical-property tests.

One 3/4-inch-thick rolled plate was given a double austenitizing treatment of 1 hour at 1650 F followed by quenching and tempering 1 hour at 350 F to provide a random-textured plate with a quenched and tempered microstructure for comparison with the thermomechanically treated plates.

Two other slabs were given special processing. One was heated to 1650 F rather than to 2100 F, cooled to 1460 F, and rolled 70 percent at 1460 F in 10 passes, followed by quenching and tempering 1 hour at 350 F. The second slab was austenitized 1 hour at 1650 F and quenched, followed by tempering for 1 hour at 1250 F to form a tempered martensite-austenite structure, rolling 70 percent in ten passes at 1250 F, and quenching and tempering 1 hour at 350 F. These two plates were used to determine the effects of soaking temperature and of tempered martensite-austenite mixtures, respectively, on the ballistic properties of thermomechanically treated steel armor.

Ballistic tests were performed on the U. S. Steel ballistic range, using 50-caliber armor-piercing projectiles at 0 degree obliquity. Ballistic plates for these tests were prepared from the 3/4-inch-thick rolled plates by surface grinding both faces to 5/8- or 1/2-inch thickness. The results are reported in terms of the velocity of the projectile which will just penetrate the plate. This velocity is referred to as the ballistic limit. Comparison of the ballistic limits of thermomechanically treated steel armor

with that of standard armor produced by conventional means was made by using existing ballistic data for standard steel armor of various thicknesses and hardnesses.

Metallographic studies included standard metallographic examination by light microscopy to determine austenite grain size, an X-ray determination of retained austenite made with a tilting-rotating attachment to minimize errors resulting from texture,[8] and an examination of the fine structure of the martensite by electron-transmission microscopy. Texture studies of specimens were performed by using MoK_α radiation. A (110) pole figure was obtained on one specimen rolled at 1365 F to determine the main texture components. The remaining specimens were examined for crystallographic texture by measuring the integrated intensities of the (110), (200), (211), (310), and (222) reflections parallel to the rolling plane. Inverse pole figures showing the pole-density distribution normal to the rolling plane were also obtained in a few cases. All the above studies were made on specimens taken from the center of the plate.

Mechanical properties of the rolled plates were measured by using standard 0.252-inch-diameter tension specimens and full-size charpy V-notch specimens. All these specimens were taken transverse to the rolling direction. Both the Young's modulus and the shear modulus in the thickness direction of the plate were measured by a pulse-echo overlap technique[9] with 5 MHz piezoelectric transducers. The density was determined by weighing specimens in air and in monobromobenzene.

Results

Determination of Critical Range

Since our earlier work[1] indicated that intercritical rolling was essential to the development of optimum ballistic properties of the present steel, the temperatures at which austenite and ferrite could coexist in this steel were carefully determined. This was done by austenitizing small specimens 1 hour at 1650 F, water-quenching to form martensite, and tempering 1 hour at various temperatures from 1050 to 1500 F. The hardness and the retained-austenite contents of these specimens were measured and are shown in Figure 1. The hardness decreased initially because of the normal softening that occurs during tempering. However, when austenite that formed during heating began to transform to martensite on quenching, the hardness increased, reaching a maximum when complete austenitization was achieved.

The hardness minimum and hardness maximum are usually taken as measures of the austenite start (A_s) and austenite finish (A_f) temperatures. However, the X-ray results show that retained austenite began to increase at 1100 F, reached a maximum at 1250 F, and then decreased to the as-quenched value at 1350 F. Such behavior is typical of high-nickel steels.[10]

Both the hardness and retained-austenite results indicated an A_f temperature of 1420 F. However, the retained-austenite results indicate that the A_s temperature is much lower than 1225 F, as

indicated by the hardness results. Calculated values of the A_s and A_f temperatures from Andrews empirical formula[11)] were 1166 and 1416 F, respectively. Although the measured and calculated A_f temperatures agree, the A_s temperature is actually between 1050 and 1100 F, much lower than that calculated.

Isothermal Transformation Diagram

A few experiments were conducted to establish the main features of the isothermal transformation diagram at 1250, 950, 850, and 650 F. Martensite start (M_s) and finish (M_f) temperatures were calculated from the empirical formula of Andrews.[11)] The A_s and A_f temperatures were taken from the retained-austenite, results as previously discussed. The isothermal transformation diagram is given in Figure 2. Ferrite forms extremely slowly at 1400 to 1200 F in this steel because of the high nickel content. Cementite also forms extremely slowly at 1400 to 1200 F and no carbide was observed even after 10 days at 1250 F. At lower temperatures (1000 F), faster transformation occurs by formation of bainite, and a bainite region typical of nickel-molybdenum steels develops.

Mechanical Properties

The mechanical properties of the plates rolled at the various temperatures are given in Table II and Figure 3. Yield and tensile strength increased with decreasing rolling temperatures. Mechanical properties of the quenched and tempered plate are given in Table II for comparison. The elastic moduli are also given in

Table II and Figure 4. These values increased continuously with decreasing rolling temperature. Elastic modulii of the quenched and tempered plate are given in Table II and Figure 4 for comparison. The measured density of the armor steel used in the elastic modulii calculations was 7.767 g/cm^3.

Preferred Orientation

A (110) pole figure for the steel rolled at 1350 F is shown in Figure 5A. The texture can be approximated by strong (112) $[\bar{1}10]$ and weaker (111) $[\bar{1}\bar{1}2]$ texture components. The inverse pole figure for the same steel is shown in Figure 5B.

The intensities of the (110), (200), (211), and (222) reflections parallel to the rolling plane were measured for the various rolling temperatures and the results are given in Figure 6. Results for both 1/2-inch and 5/8-inch plates are shown for rolling temperatures of 1520 and 1365 F. The intensities of the (112) and (111) reflections increase continuously, whereas those of the (100) and (110) reflections decrease continuously with decreasing rolling temperature. Intensities parallel to the rolling plane for the plate produced by double austenitizing, quenching, and tempering, were 0.72, 1.05, 1.09, 0.75, and 1.25 in random units for the (110), (200), (211), (310), and (222) reflections, respectively. These intensities indicate that the texture is not as near random as in the plate rolled at 1900 F (Figure 6). Evidently, some texture from the original rolling treatment was inherited even though the plate was given a double heat treatment.

Metallography

The microstructure of the quenched and tempered specimen is compared with that of a specimen thermomechanically treated at 1365 F in the light micrographs of Figures 7A and 7B. The equiaxed austenite grains developed by normal quenching and tempering were replaced by a markedly elongated austenite grain structure when the steel was rolled at low temperatures before quenching and tempering. The austenite grain size in three mutually perpendicular directions of plates rolled at different temperatures is shown in Figure 8. At high rolling temperatures the austenite grains are equiaxed, but they become more elongated as the rolling temperature decreases. The austenite grain size developed by quenching and tempering at 1650 to 1700 F was 0.06 mm. No ferrite was present in steels rolled at 1350 or 1250 F even though they were rolled in the intercritical temperature range.

The substructures of the martensite developed by rolling at high and low temperatures followed by quenching and tempering are shown in the electron-transmission micrographs of Figures 9A through 9C. The substructure of the martensite in the quenched and tempered specimen (Figure 9A) was similar to that developed by high-temperature rolling followed by quenching and tempering (Figure 9B). However, low-temperature rolling produced a finer substructure (Figure 9C).

Retained Austenite

The retained-austenite contents of the plates rolled at different temperatures were determined by X-ray diffraction by using

a tilting and rotating attachment to minimize possible effects of crystallographic texture, and these results are shown in Figure 10. Included are results for both the 1/2-inch and 5/8-inch ballistic plates. The retained-austenite content for the quenched and tempered plate is indicated for comparison. The retained-austenite content in the rolled plates was always greater than that in the quenched and tempered plate. The amount of retained austenite increased as the rolling temperature was lowered, reached a maximum at 1550 F, and then decreased at 1350 and 1250 F. The retained-austenite content of the steel heated to 1250 F and rolled at this temperature was 14.8 percent.

Ballistic Properties

The ballistic limits and merit ratings of the thermomechanically treated armor plates are given in Table III. The ballistic limit, as described earlier, is simply the penetration velocity of the projectile. The velocity merit rating is simply the ratio of the penetration velocities of the thermomechanically treated armor plate to that of standard armor plate of the same thickness. Velocity merit ratings greater than 1 indicate that the thermomechanically treated armor plate is superior to standard armor plate. The hardness merit rating is obtained in a similar manner except that a comparison of the penetration velocities is made with standard armor of the same thickness and hardness. This

requires that the penetration velocities of standard armor with a hardness of 500 BHN be corrected to the same hardness as that of the armor plate being tested using known data for the effect of hardness on the penetration velocity of standard armor. Such a hardness merit rating permits separation of differences in ballistic limit caused by factors other than hardness changes.

The effect of rolling temperature on the velocity and hardness merit ratings is shown in Figure 11. The velocity merit rating of the thermomechanically treated plates increased as the rolling temperature decreased to 1500 F, reaching a maximum value of 1.30. Below this temperature no further increase was observed. The hardness merit rating also increased as the rolling temperature was lowered to 1500 F, reaching a maximum value of 1.15, but then decreased as the rolling temperature was further lowered. Both the velocity and hardness merit ratings of the 1/2-inch armor were always higher than those of the 5/8-inch armor, even though the ratings were determined by comparison with standard armor of the same thickness as the tested plate.

The appearance of the front and back surfaces of the ballistic test plates for the quenched and tempered and thermomechanically treated conditions are shown in Figure 12. The exit hole is about 1-1/2 times larger than the entrance hole in the thermomechanically treated plate, but in the quenched and tempered plate the entrance and exit holes are the same size. The larger

exit hole was a characteristic of all the rolled plates. Cross sections of partial penetrations of plates are shown in Figure 13 for both the thermomechanically treated and quenched and tempered condition. The fracture surface of the quenched and tempered plate had an elliptsoidal shape traversing the full thickness of the plate, whereas the fracture surface of the thermomechanically treated plate was a plane parallel to the rolling plane.

Discussion

It is clear that the ballistic properties of steel armor can be improved by a thermomechanical treatment which consists of rolling at a temperature (1500 F) just above the critical temperature range followed by quenching and tempering. Such armor has ballistic properties superior to those of armor processed by simple quenching and tempering or by thermomechanical treatment at higher temperatures.

Thermomechanical treatment at lower temperatures (1350 F) within the critical range led to no further increase in the velocity merit rating of the armor and a decrease in the hardness merit rating (Figure 11). Examination of the isothermal transformation diagram (Figure 2) indicates that the low rate of austenite transformation in the present alloy steel precludes ferrite formation during intercritical rolling. This was supported by direct examination of intercritically rolled specimens in which no ferrite was detected (Figure 7B). Thus, possible further improvements in toughness because of formation of ferrite-austenite mixtures[10)]

during rolling at low temperature in the intercritical range are not practical in the present steel.

The improvement in ballistic properties as the rolling temperature is lowered (Figure 11) could be attributed to a number of causes. Rolling at lower temperatures (1) improves the mechanical properties, (2) produces a crystallographic texture in the austenite which is inherited by the martensite, (3) causes a reduction in the austenite grain size (thickness direction) and a resultant reduction in the martensite plate size, and (4) increases the amount of retained austenite.

An increase in the hardness or mechanical properties of armor plate generally results in improved ballistic performance.[12] Thus, part of the improved velocity merit rating can result from a simple increase in hardness or improvement in mechanical properties of the armor plate when it is rolled at lower temperatures. Such improvement in mechanical properties is typical of thermomechanically treated steels[13] and increases in yield strength, tensile strength, and hardness were found in the present steel after thermomechanical treatment (Figure 3). Such increases are usually attributed to an increased dislocation density or a finer martensite plate size.[13] However, when the hardness increase is corrected for by use of a hardness merit rating (Figure 11), the ballistic-property increase at lower rolling temperatures, although less, was still significant. Also, a maximum in ballistic properties occurred

at rolling temperatures of 1500 F, followed by a decrease. No such maximum was observed in the mechanical properties. Thus, the improved ballistic properties cannot result only from the improved mechanical properties.

It has recently been suggested that the toughness of ferritic-pearlitic steels is increased if a preferred orientation is developed which results in the (100) component being decreased on that plane experiencing the highest normal stress[7] since cleavage is expected to occur on the (100) plane in iron.[4] For armor plate, this plane is the original rolling plane. The texture developed in the martensite by rolling at low temperatures (Figures 5 and 6) is such that the (100) component is slightly decreased in the rolling plane and could account for part of the increased ballistic performance, since spalling (cleavage) parallel to the rolling plane would be more difficult. The preferred orientation developed is similar to that reported earlier for thermomechanically treated 4340 steel,[13] but differs from that of low-carbon steel rolled at similar temperatures, for which a strong (100) component is developed in the rolling plane.[7]

However, the crystallographic texture developed by rolling at 1550 to 1250 F has another beneficial effect since it results in an increase in the elastic moduli in the thickness direction of the plate. As shown in Figure 4, the Young's modulus (E) is increased significantly (about 3%) over that for a random quenched and tempered

plate. This increase is consistent with the increase in the (111) and a decrease in the (100) texture components.[4] Although, the stress intensity in the reflected tensile wave is proportional to $E^{1/2}$ or to the elastic wave velocity,[14] the cleavage stress is roughly proportional to E.[15] Thus, an increase in E should make cleavage more difficult and be beneficial to ballistic performance.

As can be seen from Figure 7, rolling at low temperatures causes an elongation in the austenitic grains, the primary reduction in the austenite grains occurring in the thickness direction. This extreme elongation indicates that little or no recrystallization is occurring during low-temperature rolling. The resultant martensite plate size is substantially refined, as the largest plates formed are essentially determined by the grain dimension in the thickness direction rather than in the other two directions (Figure 8). Such refinement in the martensite plate size is beneficial both to yield strength and to toughness[16] and might be expected to increase the ballistic properties. Soaking at 1650 F rather than 2100 F followed by rolling at 1460 to 1440 F led to no further increase in ballistic properties (Table III) indicating that rolling temperature rather than soaking temperature may be more important in establishing a fine austenite grain size.

The extreme elongation of the austenite grains may also have a second effect on fracture during ballistic tests. The elongated austenite grains make fracture propogation parallel to the rolling plane easier if the fracture propogates along former austenite

grain boundaries. Such "splitting" has been reported in ferritic-pearlitic line-pipe steels finished at low temperatures.[17] Although this should make the spalling process easier, such "splitting" also serves as a delamination process which will require energy and may increase notch toughness.[18] In this regard, the appearance of the entrance and exit holes of the armor plate (Figure 12) indicates that in the thermomechanically treated plates the fracture must spread as it moves into the plate, whereas this does not happen in the quenched and tempered plate. Also, cross sections of partial penetrations (Figure 13) indicate a fracture surface parallel to the rolling plane in the thermomechanically treated plate but not in the quenched and tempered plate.

The retained-austenite content of the armor plate increases as the rolling temperature is decreased, except at the lowest rolling temperatures (1365 and 1250 F), where a decrease is observed. It is well known that a finer austenite grain size leads to more retained austenite than does a coarse austenite grain size.[19] Much of the increased retained austenite at lower rolling temperature is evidently a result of the refinement of the austenite grain size. However, plastic deformation, at least for small strains, appears to accelerate the transformation of austenite to martensite.[20] Thus, at low rolling temperatures (1365 F), the effect of austenite grain refinement may be overcome by the effect of plastic deformation, and the amount of retained austenite decreased.

It is interesting to note that a maximum in the hardness merit rating of steel armor occurred at about the same temperature as the maximum in retained austenite. There is some indication that optimum combinations of strength and toughness can be developed with small amounts of retained austenite.[10,21] The amount of retained austenite to develop optimum ballistic properties in the thermomechanically treated steel appears to be about 8 percent. The plate rolled at 1250 F with the tempered martensite-austenite structure, although exhibiting a lower hardness still had a high velocity merit rating and the best hardness merit rating of any of the plates tested. An investigation of the effects of retained austenite on the ballistic properties of quenched and tempered plates appears worthwhile.

Summary and Conclusions

1. Optimum ballistic properties were obtained in a medium-carbon 5Ni-Si-Mo-Cu armor steel by rolling at temperatures (1500 F) just above the critical range followed by quenching and tempering.
2. Rolling at lower temperatures within the critical range (1365 to 1250 F) followed by quenching and tempering led to no further increase in ballistic properties. The low rate of austenite transformation in this steel precluded ferrite formation during intercritical rolling. Thus, thermomechanical treatment performed at temperatures within the critical range was essentially the same as that performed at temperatures just above the critical range.

3. Increases in the yield strength (191 to 204 ksi), tensile strength (292 to 313 ksi) and hardness (581 to 618 DPH) of the armor steel resulted when the temperature of the rolling was lowered from 1800 to 1250 F.

4. A crystallographic texture developed in the martensite as the rolling temperature of the austenite was decreased. The texture components were $(112)[\bar{1}10]$ and $(111)[\bar{1}\bar{1}2]$. Because of this texture, the (100) component parallel to the rolling plane of the plate was decreased at lower rolling temperatures.

5. Both the Young's modulus and the shear modulus in the thickness direction of the armor plate increased about 3 percent as the rolling temperature was decreased from 1800 to 1250 F.

6. Retained austenite content of the armor steel increased from 6 to a maximum value of 8 percent as the rolling temperature was decreased from 1800 to 1500 F and then decreased to 5 percent at lower rolling temperatures. Retained austenite contents of 4 percent were observed for the same armor steel in the quenched and tempered condition.

7. The austenite grain size in the thickness direction of the plate decreased markedly at lower rolling temperatures because of the marked elongation of austenite grains. This resulted in a finer martensite plate size as the rolling temperature was decreased.

8. The optimum ballistic performance resulting from thermomechanical treatment just above the critical range cannot be explained simply by the increases in strength or hardness. Rather, the improvement

also resulted from the simultaneous beneficial increase in a number of other metallurgical parameters. These included a finer martensite plate size and an increase in the amount of retained austenite, both of which have a beneficial effect on the toughness. Also, the crystallographic texture developed by low-temperature rolling makes cleavage fracture more difficult both because of a weaker (100) component in the rolling plane of the plate and because of an increase in the elastic modulus in the thickness direction of the plate.

Suggestions for Future Work

The present work suggests that small amounts of retained austenite may significantly improve ballistic properties. A more detailed study of this parameter in a number of steels is recommended.

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Table I

Chemical Composition of Armor Steel

Weight Percent

Heat	C	Mn	P	S	Si	Ni	Mo	Cu	V	N	Al
7218-8043	0.39	0.60	0.005	0.007	1.29	5.46	0.49	0.94	0.10	0.009	0.031
7464-8001	0.35	0.59	0.003	0.005	1.30	5.32	0.50	0.98	0.10	0.005	0.037
7464-8002	0.38	0.63	0.001	0.003	1.35	5.47	0.51	1.15	0.091	0.005	0.044

Table II

Mechanical Properties of Thermomechanically Treated Steel Armor

Soaking Temperature	Thermomechanical Treatment	Yield Strength, 10^3 psi	Tensile Strength, 10^3 psi	Reduction in Area, pct	Elongation, pct	Charpy V-Notch Energy, ft-lb	Hardness, DPH	Youngs Modulus, 10^6 psi	Shear Modulus, 10^6 psi
				Thermomechanically Treated					
2100 F	70% reduction at 1900-1750 F/WQ/1h 350 F	191±4	292±2	47.7	15	19	581	28.52	11.12
2100 F	70% reduction at 1700-1645 F/WQ/1h 350 F	203±2	297±1	-	-	18	576	29.13	11.23
2100 F	70% reduction at 1530-1500 F/WQ/1h 350 F	205±3	304±1	43	13	18	571	29.41	11.41
2100 F	70% reduction at 1400-1350 F/WQ/1h 350 F	204±1	313	38	12	18	576	29.75	11.51
2100 F	70% reduction at 1260-1240 F/WQ/1h 350 F	226	313	34	13	23	618	29.27	11.31

(Continued)

Table II (Continued)

Soaking Temperature	Thermomechanical Treatment	Yield Strength, 10^3 psi	Tensile Strength, 10^3 psi	Reduction in Area, pct	Elongation, pct	Charpy V-Notch Energy, ft-lb	Hardness, DPH	Youngs Modulus, 10^6 psi	Shear Modulus, 10^6 psi
	Thermomechanically Treated (Continued)								
1650 F	70% reduction at 1460-1440 F/WQ/1h 350 F	217	285	31	10	17	601	29.10	11.25
1250 F	70% reduction at 1250-1150 F/WQ/1h 350 F*	197	289	24	15	15	549	27.81	10.60
	Quenched and Tempered								
	1h 1650 F/WQ/1h 350 F**	214±1	287±1	33	10	16	570	28.92	11.09

* Initially austenitized 1 H 1655 F/WQ.

** Doubly austenitized, quenched and tempered.

WQ - Water quench.

Table III

Ballistic Properties of Thermomechanically Treated Steel Armor

Soaking Temperature	Thermomechanical Treatment	Thickness, in.	Hardness, BHN*	Observed Ballistic Limit, ft/s**	Merit Rating*** Velocity Merit Rating	Merit Rating*** Hardness Merit Rating
			Thermomechanically Treated			
2100 F	70% reduction at 1900-1750 F/WQ/1h 350 F	0.620 0.502	545 532	2195 1892	1.07 1.06	0.945 0.967
2100 F	70% reduction at 1700-1645 F/WQ/1h 350 F	0.621	541	2341	1.14	1.020
2100 F	70% reduction at 1530-1500 F/WQ/1h 350 F	0.618 0.502	536 542	2401 2337	1.18 1.30	1.07 1.15
2100 F	70% reduction at 1400-1350 F/WQ/1h 350 F	0.619 0.498	541 560	2411 2307	1.18 1.29	1.04 1.07
2100 F	70% reduction at 1260-1240 F/WQ/1h 350 F	0.508	582	2381	1.30	1.10
1650 F	70% reduction at 1460-1420 F/WQ/1h 350 F	0.508	564	2341	1.30	1.17
1250 F	70% reduction at 1250-1150 F/WQ/1h 350 F ****	0.505	515	2218	1.23	1.28

(Continued)

Table III (Continued)

Soaking Temperature	Thermomechanically Treatment	Thickness in.	Hardness BHN*	Observed Ballistic Limit, ft/s**	Merit Rating***	
					Velocity Merit Rating	Hardness Merit Rating
			Quenched and Tempered			
	1h 1650 F/WQ/1h 350 F ****	0.500	535	1994	1.12	0.99

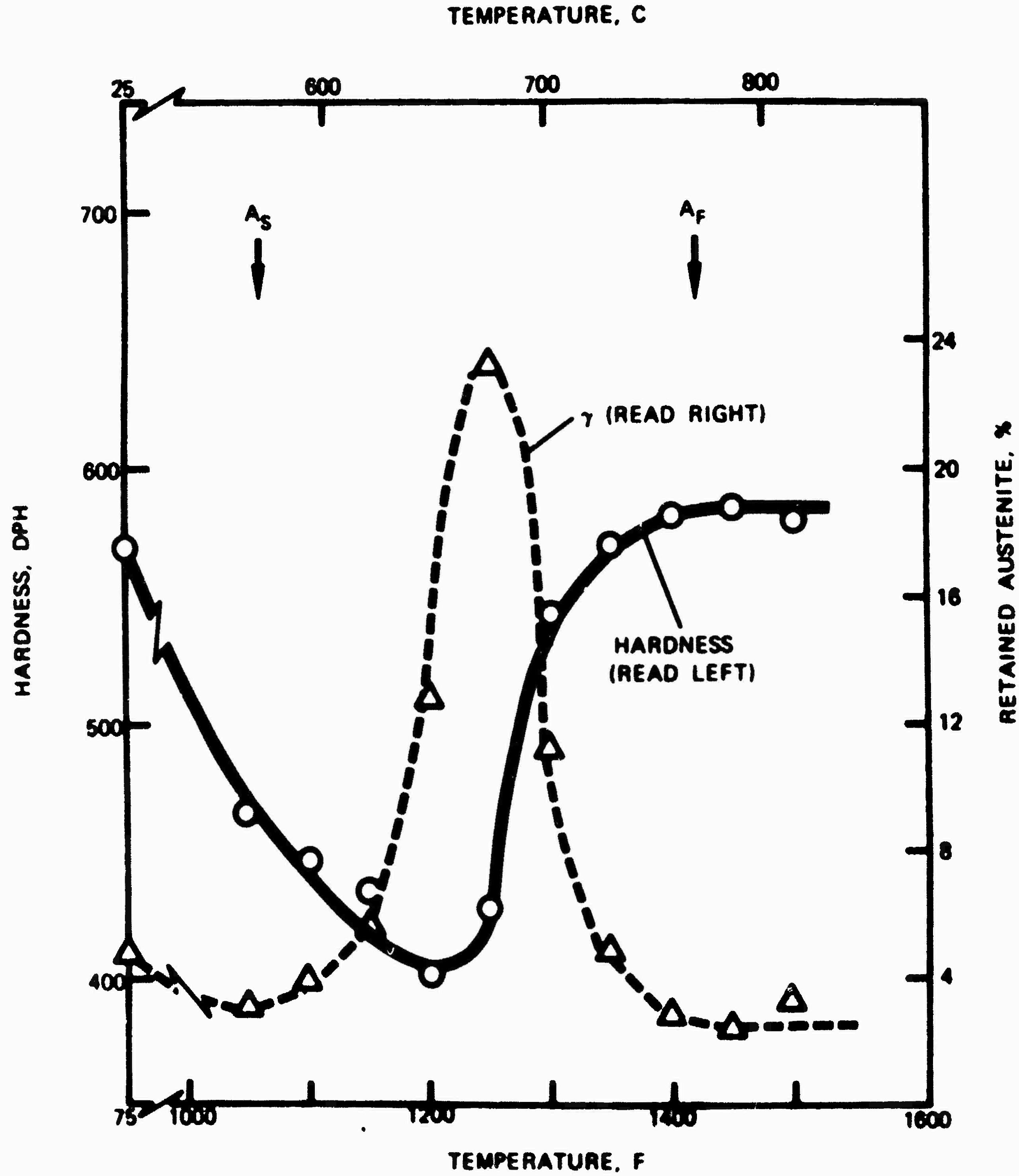
* ASTM 140 conversion from DPH.

** 0.50 caliber APM2 projectiles, 0° obliquity.

*** Compared with standard quenched and tempered armor of similar hardness and thickness.

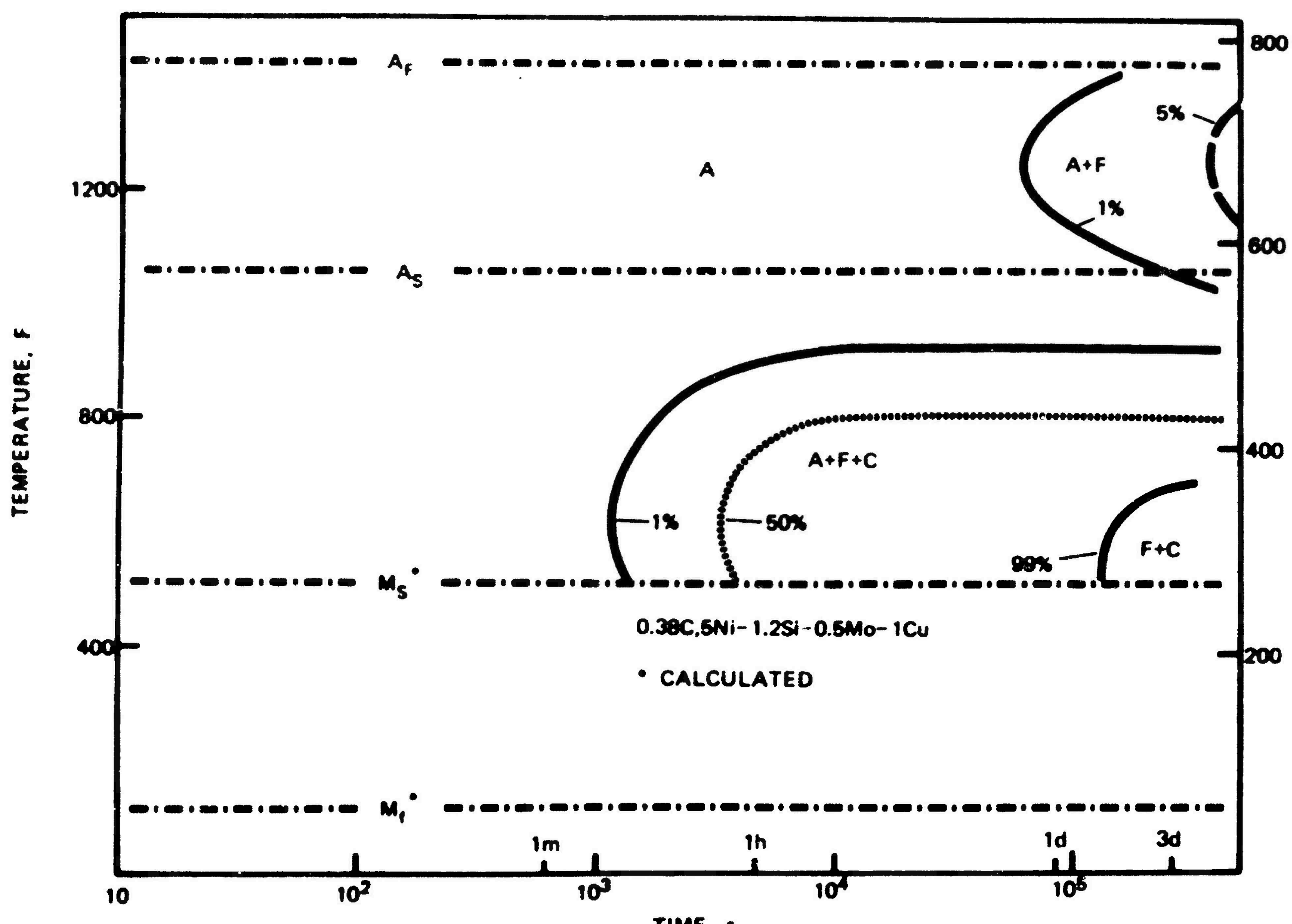
**** Initially austenitized 1 h 1650 F/WQ.

WQ - Water quench.



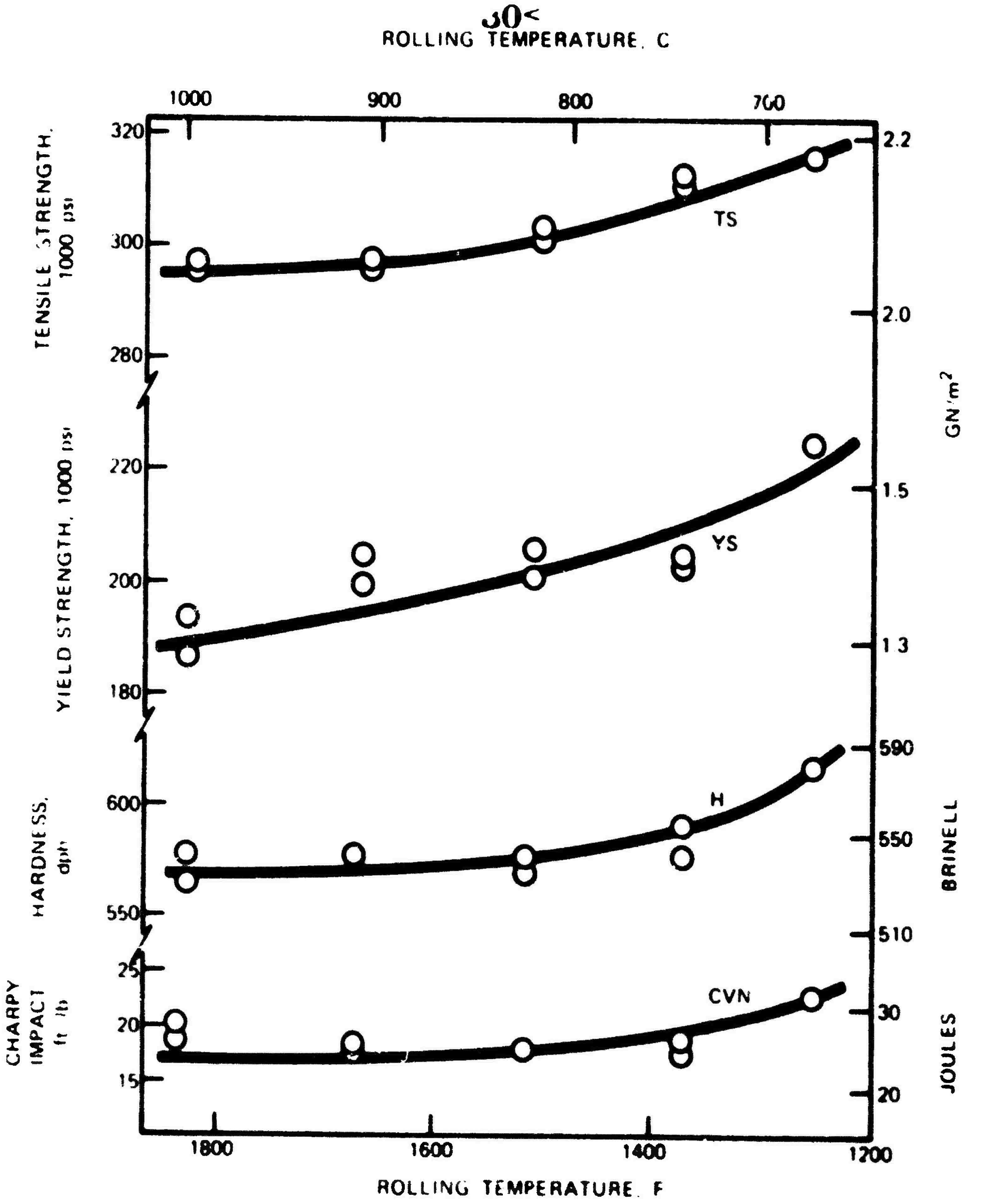
AUSTENITE START AND FINISH TEMPERATURES OF ARMOR STEEL

Figure 1



ISOTHERMAL TRANSFORMATION DIAGRAM OF ARMOR STEEL

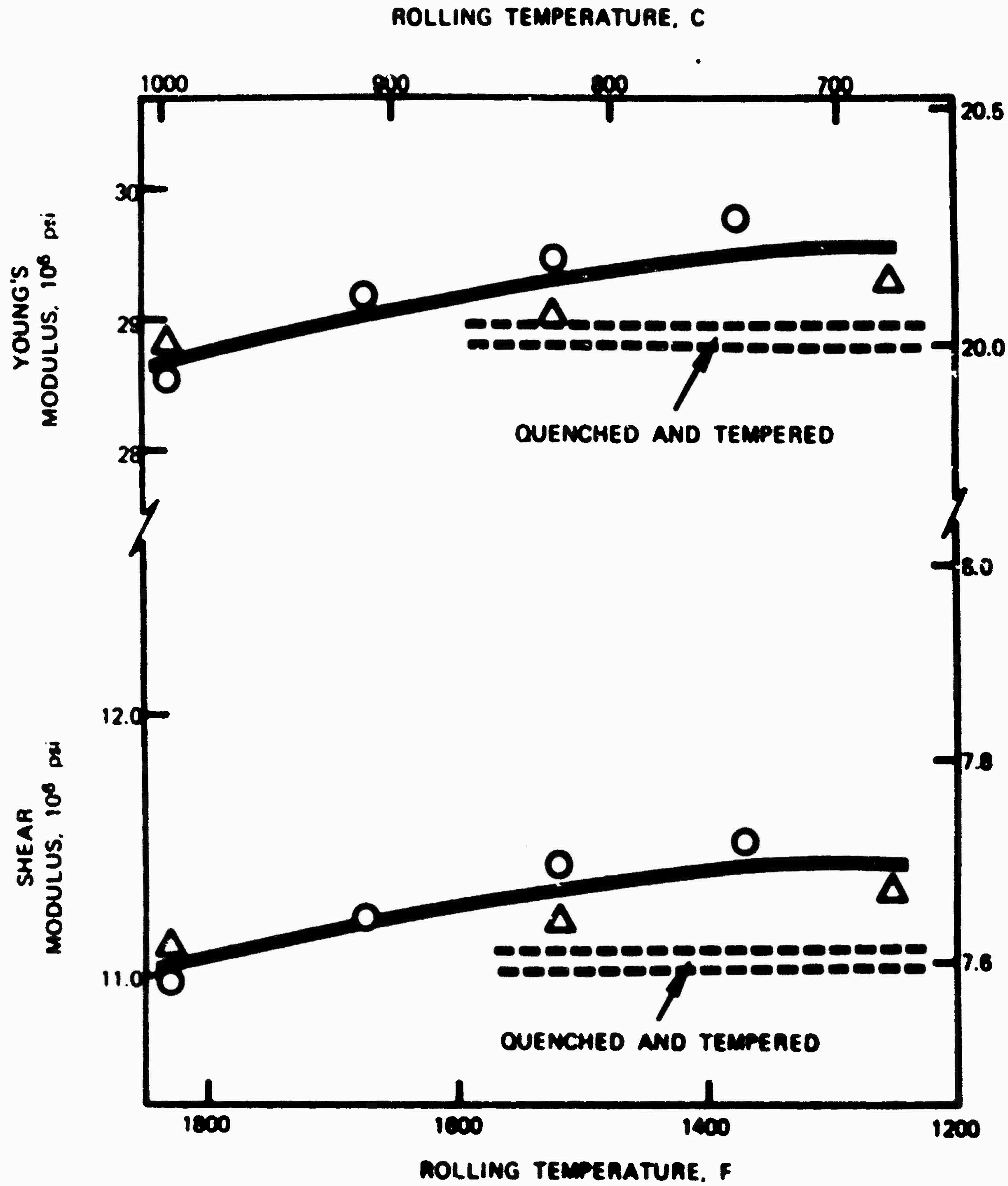
Figure 2



EFFECT OF ROLLING TEMPERATURE ON MECHANICAL PROPERTIES OF THERMOMECHANICALLY TREATED ARMOR STEEL

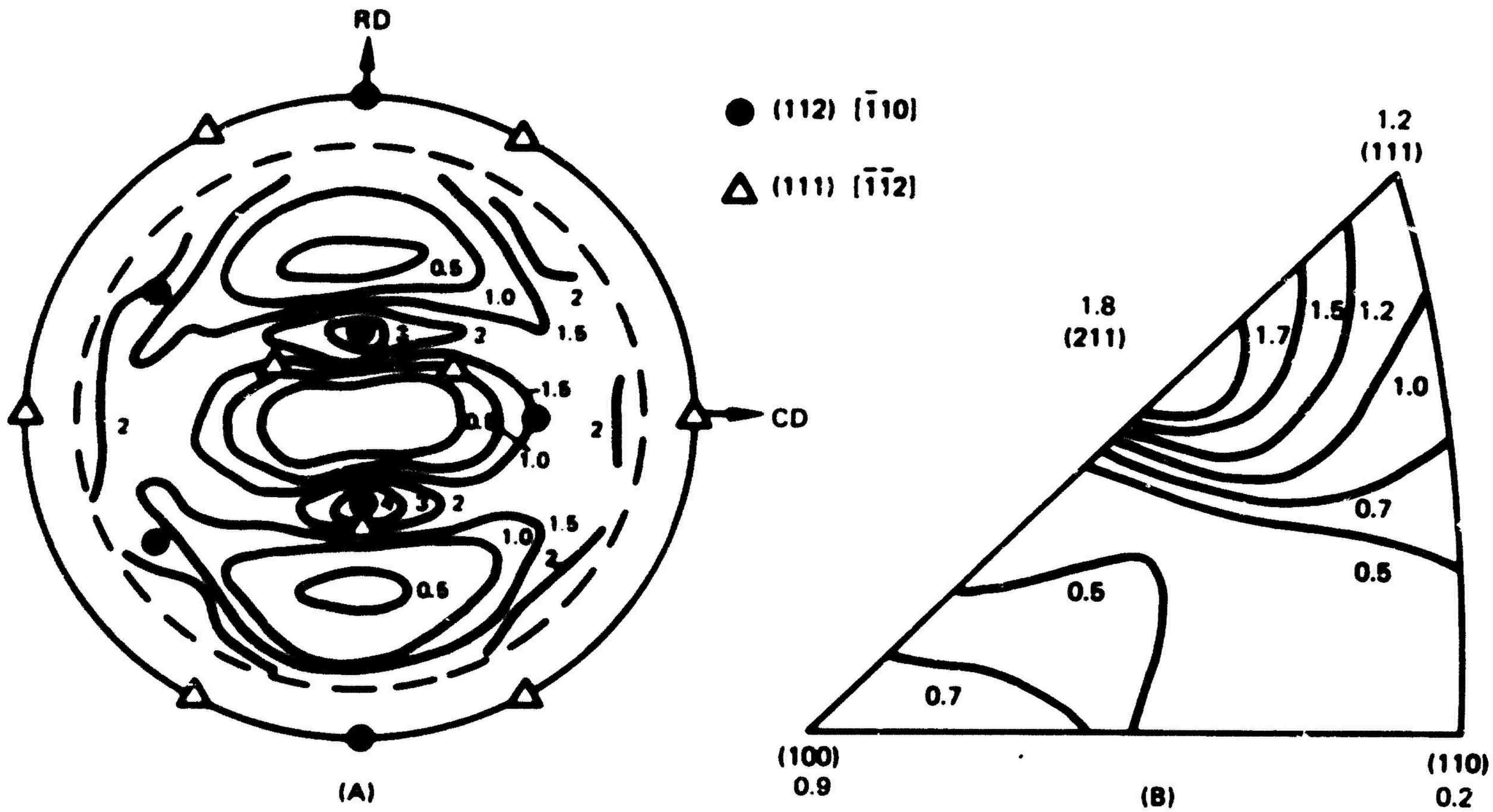
Figure 3

31<



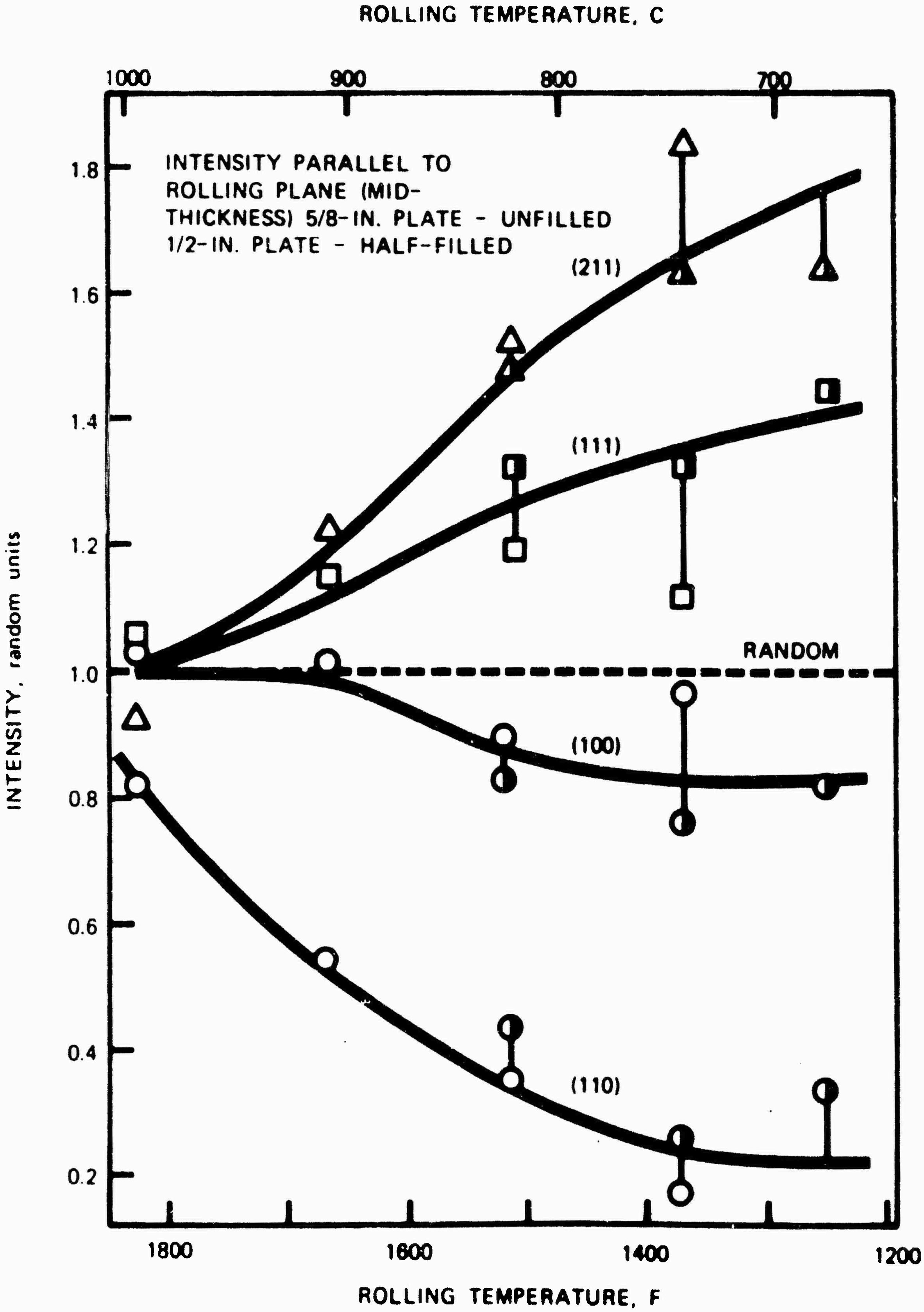
EFFECT OF ROLLING TEMPERATURE ON THROUGH-THICKNESS ELASTIC MODULI OF THERMOMECHANICALLY TREATED ARMOR STEEL

Figure 4



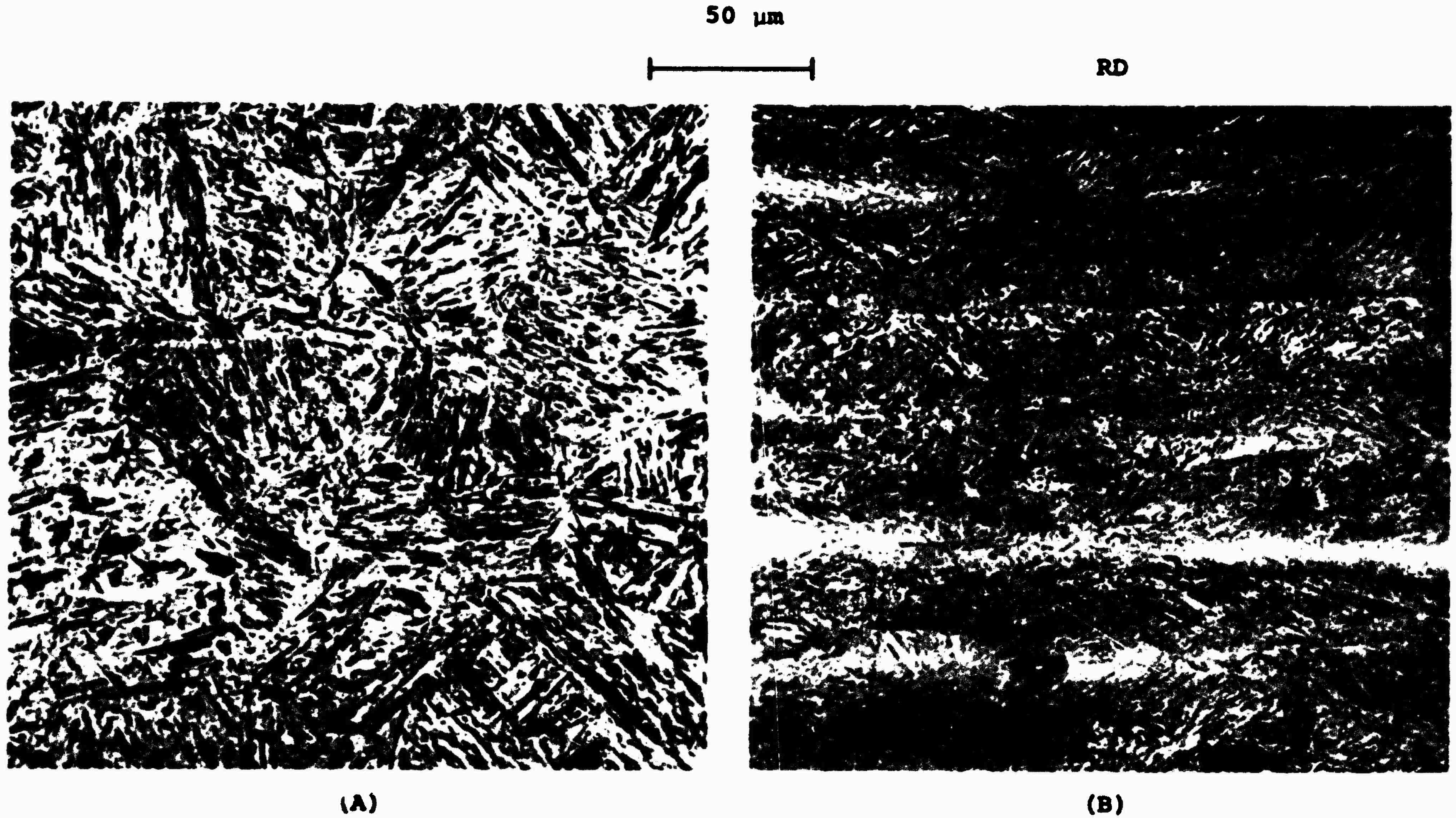
PREFERRED ORIENTATION IN ARMOR STEEL ROLLED AT 1365 F, QUENCHED AND TEMPERED. (A) (110) POLE FIGURE, (B) INVERSE POLE FIGURE.

Figure 5



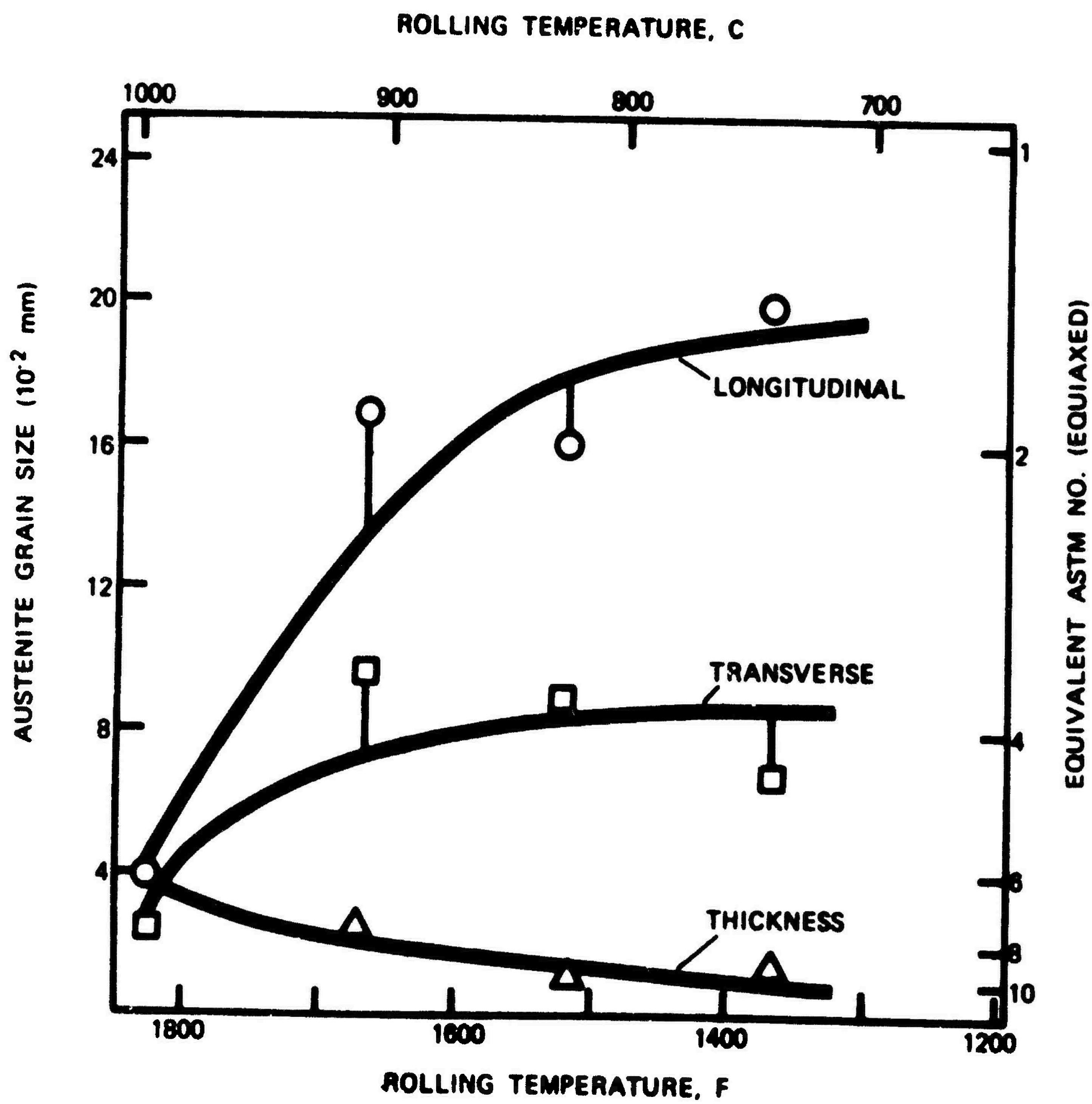
EFFECT OF ROLLING TEMPERATURE ON PREFERRED ORIENTATION OF THERMOMECHANICALLY TREATED ARMOR STEEL

Figure 6



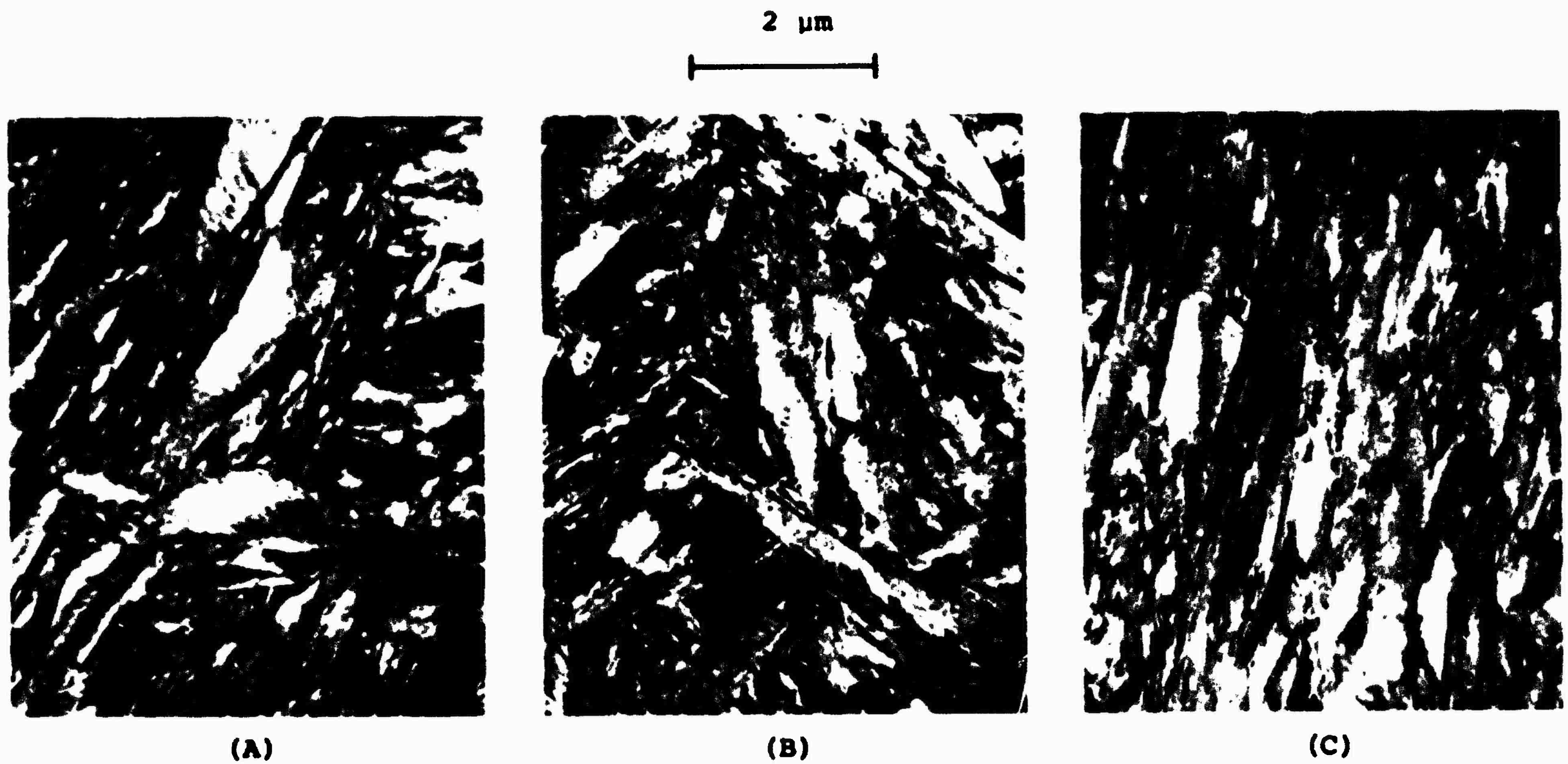
Microstructures of armor steel (A) quenched and tempered, (B) rolled 70% at 1365 F.

Figure 7



EFFECT OF ROLLING TEMPERATURE ON AUSTENITE GRAIN SIZE OF THERMOMECHANICALLY TREATED ARMOR STEEL

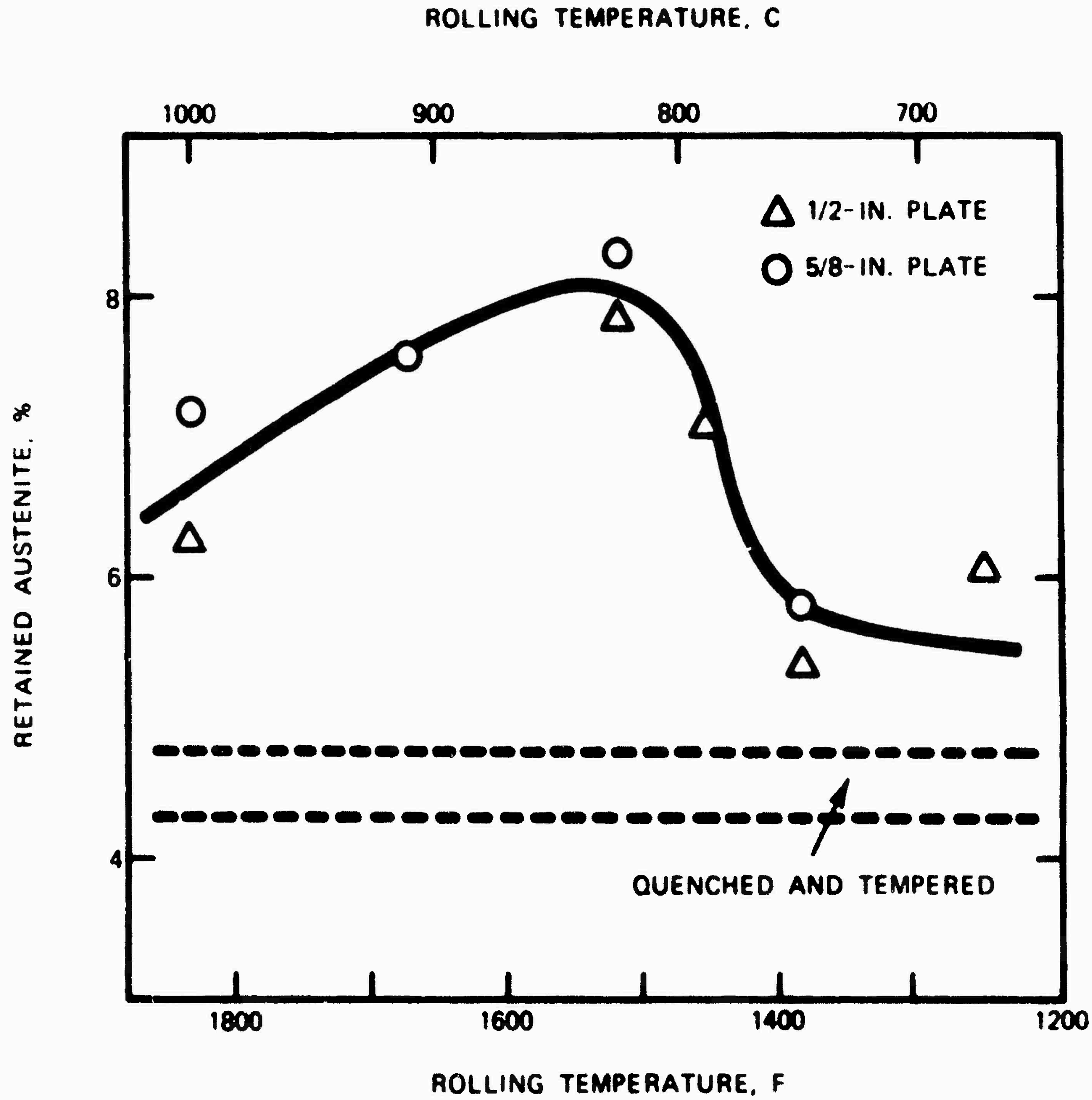
Figure 8



Substructures of armor steel (A) quenched and tempered, (B) rolled 70% at 1825 F, (C) rolled 70% at 1365 F.

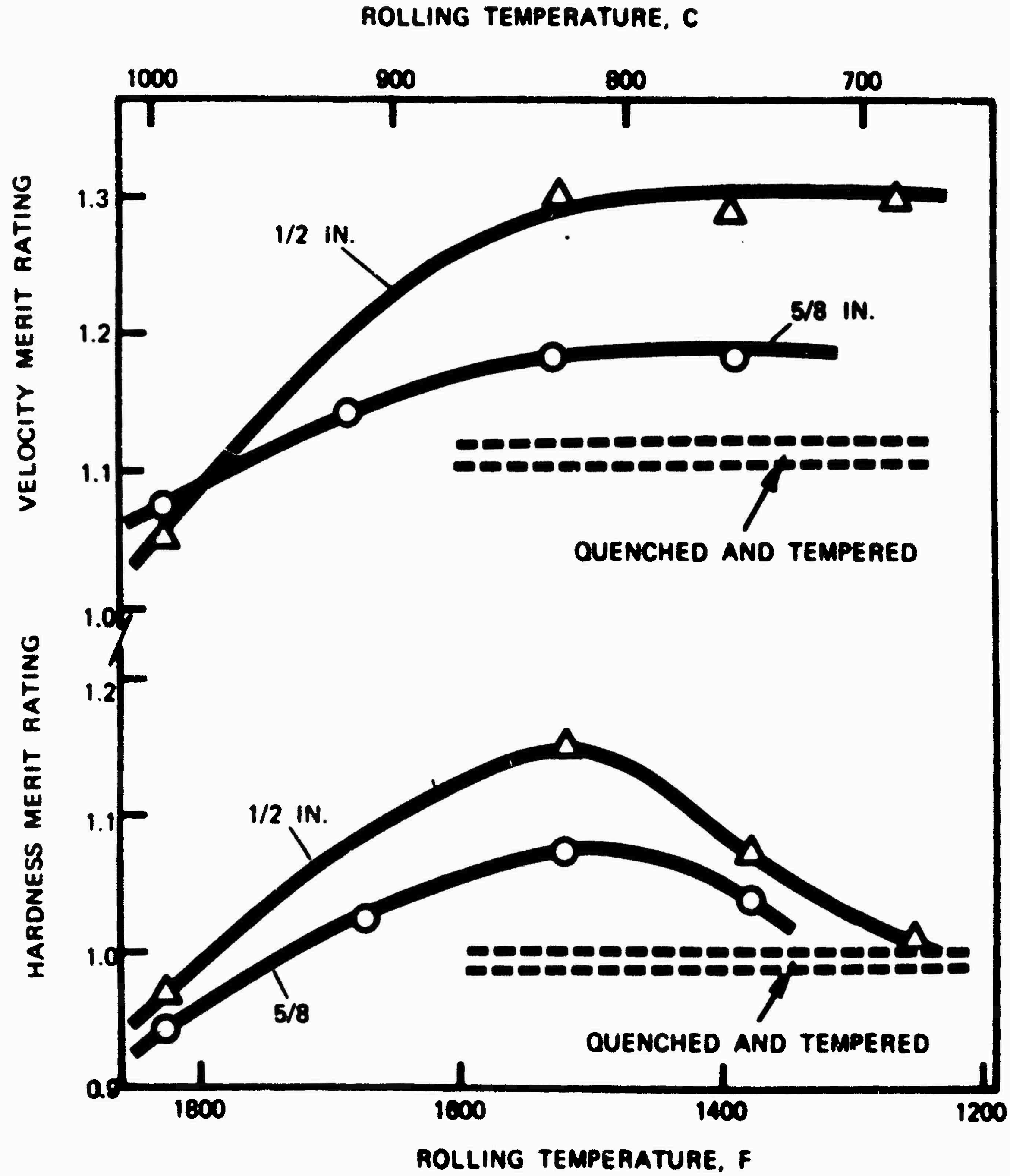
Figure 9

ل7<



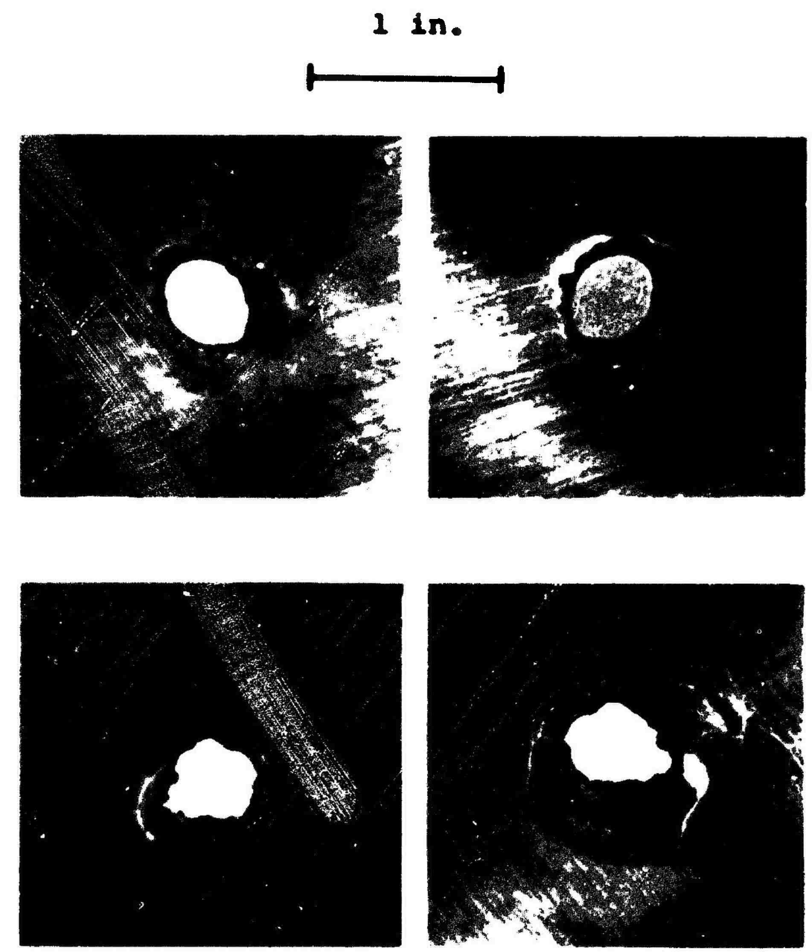
EFFECT OF ROLLING TEMPERATURE ON RETAINED AUSTENITE CONTENT OF THERMOMECHANICALLY TREATED ARMOR STEEL

Figure 10



EFFECT OF ROLLING TEMPERATURE ON BALLISTIC PROPERTIES OF THERMO-MECHANICALLY TREATED 1/2- AND 5/8-INCH-THICK ARMOR PLATE

Figure 11



(A)

(B)

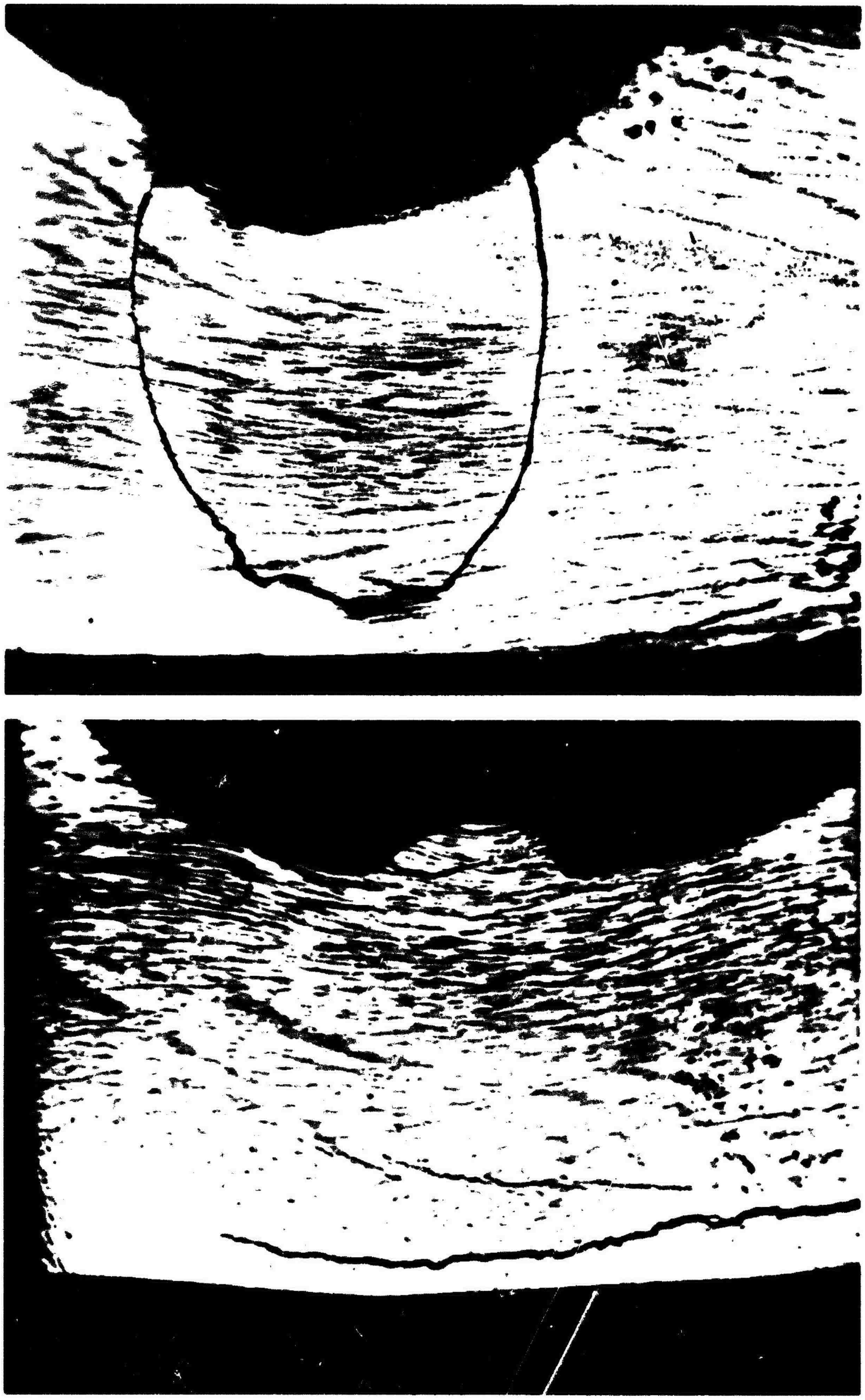
Entrance hole

Exit hole

Appearance of entrance and exit holes of ballistically tested armor plate (A) quenched and tempered (1/2 in. plate, 2190 ft/s), (B) rolled 70% at 1365 F (1/2 in. plate, 2420 ft/s).

Figure 12

1/8 in.



(A)

(B)

Partial penetrations of armor plate (A) quenched and tempered (2100 ft/s), (B) rolled 70% at 1250 F (2330 ft/s).

Figure 13